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Short fatigue crack propagation during low-cycle, high cycle and very-high-cycle fatigue of duplex steel – An unified approach

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ABSTRACT

The present paper reviews experimental results on the fatigue damage of austenitic–ferritic duplex steel under various load levels ranging from LCF to VHCF, placing the focus towards the relationship between the crystallographic orientation of individual grains and grain patches that exhibit slip band formation, fatigue crack initiation and growth. A combination between fatigue testing of electropolished specimens and analytical electron microscopy (SEM/EBSD, TEM) revealed that under LCF loading conditions almost all the ferrite and the austenite grains showed plasticity, while under HCF and VHCF loading conditions, slip band formation was limited to the softer austenite grains and a low plastic activity is observed in the ferrite. Once being formed, the bands generate high stress concentrations, where they impinge the α–γ phase boundaries, eventually, leading to the crack initiation. This is discussed by applying a numerical simulation approach based on the finite-element (FEM) and the boundary-element (BEM) method.

1. Introduction

Duplex stainless steels (DSS) are increasingly used for industrial applications that involve fatigue loading in corrosive environments. Due to the high Cr and Mo content DSS exhibit a superior resistance to stress-corrosion cracking. The mechanical properties can be tailored for the two phases, bcc ferrite (α) and fcc austenite (γ), separately by heat treatment (475 °C embrittlement) or nitrogen alloying [1].

Fatigue damage generally initiates at plastic strain concentration sites. Under low-cycle fatigue (LCF) loading conditions, plastic deformation involves large areas of a structure and causes immediate multiple-site crack initiation. Previous studies in the LCF regime have revealed that when the material is in the as-received form, all the ferrite and the austenite grains showed plastic deformation [2].

The lower critical shear stress and the planar slip character of the dislocations produce a homogeneous slip band distribution earlier in the austenite than in the ferrite. The associated dislocation structure belongs to dislocation bands on one or two {1 1 1} slip planes. In the ferrite, dislocations present the typical wavy slip character and the slip markings in the specimen surface are composed of all kind of slip features, i.e., slip lines, band-like extrusions, curvilinear cord-like extrusion and highly rugged areas. Strain localizes in these rugged zones, which promote crack initiation [3,4]. According to other authors, microcracks may nucleate in an intergranular manner in the ferrite areas, which is in agreement with the analysis of Vogt et al. [4], or along persistent slip bands in those grains where single slip is more favourable. Microcracks propagate in an oscillating way until their coalescence [5,6]. In particular in the case of high mean stresses, crack coalescence strongly contributes to short crack propagation, which can be attributed to a high crack density (cf. [7]). Finally, a large portion of the fatigue life is determined by long fatigue crack propagation (technical cracks, cf. Fig. 1a), which can be assessed by, e.g., the empirical Paris law (for an overview see [8]).

Contrary to that, high-cycle (HCF) and very-high-cycle fatigue (VHCF) damage occurs locally, when the applied load is macroscopically elastic. Second phases of different mechanical properties as well as crystallographically misoriented grains give raise to stress concentration that leads to the development and growth of slip bands during cycling [9,10]. These bands may be blocked by or transmitted through grain and phase boundaries. Due to partial irreversibility of cyclic plasticity along single slip bands (cf. Section 4), vacancy-type annihilation of dislocations moving back and forth at the slip band can be considered as infinitesimal crack advance. Arrest and overcoming grain and phase boundaries by the plastic zone and the crack, respectively, give rise to an oscillating fatigue crack propagation rate. In the case of VHCF, this oscillation may prevail throughout the complete fatigue life [10,11]. If the microstructural barriers are strong enough, all the initiated cracks

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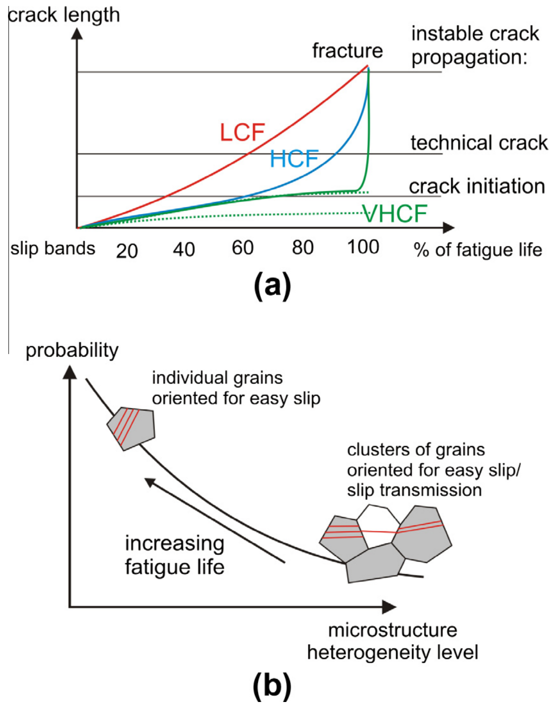

Fig. 1. Schematic representation of (a) the damage development (crack length) during LCF, HCF and VHCF, and (b) the probability of the occurrence of microstructure heterogeneities as a function of the heterogeneity level (cf. [12]).

are blocked, and the fatigue limit is reached. Accordingly, the resistance to fatigue damage for low stress amplitudes (HCF and VHCF) becomes strongly dependent on the microstructure heterogeneity level and the probability of the occurrence of critical heterogeneities [12], the latter being dependent on the specimen or component size and the production process, respectively. This is shown schematically in Fig. 1b. In the case of duplex steel, the response to fatigue loading depends on the initial mechanical conditions (stress or strain range, strain rate, etc.) and on the mechanical strength of the two phases [1,8,13]. In earlier work it was found that cyclic plasticity is partitioned mainly to the softer austenite phase leading to the occurrence of elastic residual stresses in the ferrite grains during HCF and VHCF or the mutual interaction of the phases to accommodate plastic deformation during LCF. Therefore, slip bands and microstructural crack formation cause stress relief and the onset of the fatigue damage process.

The present paper summarizes previous results on HCF and VHCF in the homogenized DSS and adds new results of LCF and HCF experiments varying the strength of the ferritic phase by an embrittlement heat treatment. As well, it will be outlined how the apparently different damage mechanisms can be assessed by a uniform short crack modelling concept.

2. Experimental

The duplex stainless steel used in this study, with the chemical composition according to Table 1, was given a grain-coarsening heat treatment in order to ease 2D and 3D measurements of the crystallographic orientation distribution. The strength of the ferrite phase was varied by means of an embrittlement heat treatment causing the formation of Cr-rich α' precipitates within the ferrite grains by spinodal decomposition [1]. The heat treatment parameters and the corresponding yield stress, hardness and grain-size values are represented in Table 1.

The two-phase microstructure (phase fractions about 50% α and γ, resp.) is shown in Fig. 2a, in a cross section normal to the rolling direction. Mechanical testing was carried out on various mechanical conditions in a wide fatigue range. HCF and VHCF tests were performed in resonant fatigue testing (RUMUL Testtronic, 40–260 Hz), servohydraulic (MTS 810, 1000 Hz) and ultrasonic (BOKU 20 kHz) testing systems using different specimen geometries, mainly under stress control and a stress ratio of $R = -1$. Fig. 2b and c show the shallow-notched specimen design that allows limitation of crack formation within an electropolished notch area (notch factor $\beta \approx 1.1$). In-situ observations with a long-distance QUESTAR optical microscope coupled to a digital camera at the testing machine were made. As well, a discontinuous evaluation of fatigue damage at the surface by scanning electron microscopy in combination with automated electron back scatter diffraction (EBSD) were carried out [14,15]. For LCF tests an electromechanical Instron machine mod.1362 was used. Testing was performed under plastic-strain control with a fully reversed triangular wave at total strain rate of $2 \times 0^{-3}\ s^{-1}$. In order to characterize the surface damage, the test was stopped (at 80% of σ_{Max}) periodically. Fig. 2c shows the shallow-notched specimen design that allows crack initiation in a limited electropolished area. The flat part of the notch for LCF was systematically explored during the test using a high resolution CCD camera JAI mod. CM-140MCL with a 50× objective and a 12× ultra zoom device mounted on the fatigue test machine.

3. Results

During LCF in the embrittled condition, the austenite phase presents – as in the as-received condition – homogeneous distribution of planar slip markings, while the ferrite exhibits first straight slip lines, which turn into slip bands during ongoing cycling. One characteristic feature of these bands is that they are heterogeneously distributed on the specimen surface as is illustrated in Fig. 3a and b. In fact, the ferrite presents planar slip- or less wavy slip-character due to spinodal decomposition; dislocations are more confined to their glide planes not being able to cross slip as usually occurs in a bcc structure [16]. This fact was proven in earlier papers by Gironés et al. and Armas et al. [17,18]. During fatigue, the back and forth dislocation motion diminishes the amplitude of the spinodal decomposition [19], which favours the subsequent dislocation motion and the generation of localized deformation regions

Table 1
Chemical composition (in wt%) and heat treatment parameters of the duplex steel (German designation 1.4462) used in this study.

Fe	C	Cr	Ni	Mo	Mn	N	P	S
Bal.	0.02	21.9	5.6	3.1	1.8	0.19	0.023	0.002

Homogenized condition + grain coarsening	1250 °C (4 h), followed by slow cooling to 1050 °C and, eventually, water quenching Ferrite: grain size 33 μm, hardness 280 HV Austenite: grain size 46 μm, hardness 260 HV Yield stress: 550 MPa
Embrittled condition + grain coarsening	1250 °C (4 h) followed by slow cooling to 1050 °C and, eventually, water quenching 475 °C (100 h) Ferrite hardness 465 HV Yield stress: 830 MPa